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(54) METHOD OF HEAT TREATING A COLD ROLLED STEEL STRIP

VERFAHREN ZUR WÄRMEBEHANDLUNG EINES KALTGEWALZTEN STAHLBANDES

PROCÉDÉ DE TRAITEMENT THERMIQUE D’UNE BANDE D’ACIER LAMINÉE À FROID

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Description

[0001] The present invention relates to a method of heat treating a high strength cold rolled steel strip.

[0002] In the art various types of cold rolled steels and manufacturing processes have been proposed for meeting the requirements in automotive applications. E.g. extra low carbon steel is used in automotive steel strip in view of its formability. This steel type shows a tensile strength in the range of 280-380 MPa.

[0003] HSLA (high strength low alloy) steels contain microalloying elements. They are hardened by a combination of precipitation and grain refining.

[0004] Advanced high strength steels (AHSS), such as dual phase (DP) steels and transformation induced plasticity (TRIP) steels, are currently typical of high-ductility, highstrength steels that are used in the automotive manufacturing industry. In DP steels the presence of martensite within a ferrite matrix enables a tensile strength higher than 450 MPa combined with good cold formability to be obtained.

[0005] To achieve simultaneously a high yield strength/tensile strength ratio and an even higher tensile strength, i.e., above 800 MPa, steels having complex (CP) microstructures including ferrite, bainite, martensite and/or retained austenite have been developed. However, due to the difference of the deformation capabilities between the ferrite, bainite or martensite structures and the retained austenite structure, these steels are generally inferior in stretch flange formability. Therefore their use is limited to automobile parts which do not require high formability.

[0006] TRIP type tempered martensitic steel (Q&P steel through quench and partitioning) that consists of tempered martensite as the matrix phase and residual austenite, and TRIP type bainitic ferrite steel (TBF steel through austempering) that consists of bainitic ferrite as the matrix phase and residual austenite have advantages such as the capability to provide high strength due to the hard tempered martensite and/or bainitic ferrite structure, and the capability to show outstanding elongation because the matrix is carbide-free, and fine residual austenite grains can be easily formed at the boundary of lath-shaped bainitic ferrite in the bainitic ferrite structure. Therefore, carbide-free bainitic ferrite or tempered martensitic steels are expected to achieve good stretch flangeability due to their uniform fine lath structure. The heterogeneities of hardness due to the presence of only a small amount of martensite in these microstructures will allow these steel types to achieve good deep drawability.

[0007] However, due to the limitations of current continuous production lines, the expected beneficial combination of the strength and ductility properties could not be obtained with current available steel recipes. These limitations comprise inter alia that the reheating furnace of current facilities of continuous annealing (CA) and continuous galvanizing (CG) lines are often only suitable for subjecting the steel strips to an intercritical or recrystallization heat treatment. For example, in some current annealing lines the maximum annealing temperature is limited to 890 °C. Furthermore the cooling rates in the current CA/CG lines are limited within a fixed range. Also the available overageing time for many CA/CG lines is limited, e.g. this time span is less than about 160 seconds, which puts serious time limits to completion of any desired transformation during overageing.

[0008] E.g. WO2013/144373A1 has disclosed a cold rolled TRIP steel with a matrix of polygonal ferrite having a specific composition comprising chromium and a particular microstructure and having a tensile strength of at least 780 MPa, which is said to allow production thereof in a conventional industrial annealing line having an overageing/austempering section. That is to say for a relatively high overageing/austempering temperature the austempering time could be less than 200 seconds.

[0009] EP2831296B1 and EP2831299 have disclosed TBF steels, having a tensile strength of at least 980 MPa which could also be produced on a conventional production line. However, the preferred overageing/austempering times being 280-320 seconds, are too long to allow production on quite a number of conventional production lines. In other words, the bainitic transformation kinetics is too slow to complete the bainitic transformation in the limited time span in the overaging section to obtain the required microstructure in a conventional production line. WO 2018/115 936 discloses a tempered and coated steel sheet having excellent mechanical properties suitable for use in manufacturing of vehicles.

[0010] An object of the disclosure is to provide a cold rolled steel strip having a desired combination of high tensile strength and excellent ductility, such as yield strength (YS) ≥ 550 MPa, tensile strength (TS) ≥ 980 MPa, total elongation (TE) ≥ 13%, hole expansion capacity (HEC) ≥ 20% and bending angle (BA) ≥ 80°, in particular a steel strip for use in automotive applications, or a suitable alternative.

[0011] A further object of the invention is to provide a method for heat treating a cold rolled steel strip for obtaining the desired combination of properties as mentioned above, in particular a heat treatment that can be carried out using existing production lines, or a suitable alternative.

[0012] Another object of the invention is to provide a high silicon cold rolled steel strip having a desired combination of properties, which can be made on conventional industrial production lines.

[0013] Yet another object of the invention is to provide a steel composition for a high strength cold rolled steel strip and heat treatment thereof allowing to complete the bainitic transformation in a conventional production line in order to obtain a desired microstructure.

[0014] In view thereof the invention provides a method of heat treating a cold rolled steel strip according to claim 1.

[0015] The method of the invention allows producing a cold rolled steel strip having a specific composition and microstructure and a combination of properties desirable for automotive parts requiring high strength, formability and weldability.

[0016] The invention solves the problem of the slow bainitic transformation kinetics by introducing a suitable amount of pro-eutectoid ferrite and controlling the morphology of it, by obtaining fine grains of the austenite through controlling the top annealing temperature and time, and by using a modified quenching and partitioning process in a production line.

[0017] This method according to the invention can be performed using existing continuous annealing and galvanizing lines within the limitations regarding top temperature in the annealing section, cooling rate ranges and overageing time window at production speeds that are typical to these production lines.

[0018] The cold rolled steel strip may be Zn coated e.g. by hot dip galvanizing or electrogalvanizing. A hot dip galvanizing step can be integrated easily in the heat treatment according to the invention.

[0019] The terms used to describe the critical transformation temperatures of a steel are given as follows, as well known to a person skilled in the art.

[0020] Ae3: Temperature at which transformation of ferrite into austenite or austenite into ferrite occurs under equilibrium conditions.

[0021] Ac3: Temperature at which, during heating, transformation of the ferrite into austenite ends. Ac3 is usually higher than Ae3, but tends towards Ae3 as the heating rate tends to zero. In this invention, Ac3 is measured at a heating rate of 3 °C/s.

[0022] Ar3: Temperature at which austenite begins to transform to ferrite during cooling.

[0023] Bs: Temperature at which, during cooling, transformation of the austenite into bainite starts. Bn: Nose temperature of the bainitic transformation in the time-temperature transformation (TTT) curve of a steel, at which transformation of the austenite into bainite has the fastest kinetics.

[0024] Ms: Temperature at which, during cooling, transformation of the austenite into martensite starts.

[0025] Mf: Temperature at which, during cooling, transformation of the austenite into martensite ends. A practical problem with Mf is that the martensite fraction during cooling approaches the maximum achievable amount only asymptotically, meaning that it takes very long for the last martensite to form. For practical reasons and in the context of this invention, Mf is therefore taken as the temperature at which 90% of the maximum achievable amount of martensite has been formed.

[0026] These critical phase transformation temperatures can be determined by dilatometer experiments. Alternatively, the Ac3, Bs, Bn and Ms points of the steel according to the invention can be calculated beforehand based on its composition using available software such as JmatPro or using the following empirical formulae:

$$\mathrm{Ac3\ (°C)} = 942 - 260\mathrm{C} + 35\mathrm{Si} - 35\mathrm{Mn} + 125\mathrm{Al} - 11\mathrm{Cr} - 14\mathrm{Cu}$$

$$\mathrm{Bs\ (°C)} = 839 - 86\mathrm{Mn} - 23\mathrm{Si} - 67\mathrm{Cr} + 35\sqrt{\mathrm{Al}} - 270(1-\exp(-1.33\mathrm{C}))$$

$$\mathrm{Ms\ (°C)} = 539 - 423\mathrm{C} - 30.4\mathrm{Mn} - 7.5\mathrm{Si} + 30\mathrm{Al}$$

[0027] In these formulae, the component X of the steel composition is represented in wt.%.

[0028] In this specification all temperatures are represented in degrees Celsius, all compositions are given in weight percentage (wt.%) and all the microstructures are given in volume percentage (vol.%) except where explicitly indicated otherwise.

[0029] In the attached drawing:

Fig. 1 is an EBSD map showing characteristics of the bainitic ferrite microstructures of a low temperature bainitic ferrite and/or partitioned martensite (Fig. 1a) and a high temperature bainitic ferrite (Fig. 1b) respectively.
Fig. 2 is a histogram of misorientation angle of a low temperature bainitic ferrite and a high temperature bainitic ferrite.
Fig. 3 is a diagram showing a generally applicable time vs temperature profile of an embodiment of the method according to the invention.

[0030] Below an explanation of the composition, the method steps and microstructure according to the invention is presented.

Composition:

Carbon: 0.15 - 0.28%

[0031] A sufficient amount of carbon is required for strength and stabilizing the retained austenite, the latter offering the TRIP effect. In view thereof the amount of carbon is higher than 0.15%, preferably higher than 0.17%. Increasing the carbon content results in an increase of the steel strength, the amount of retained austenite and the carbon content in the retained austenite. However, weldability of the steel is significantly reduced as the carbon content is higher than 0.25%. For applications that require welding, the carbon content is preferably 0.15 - 0.25%, more preferably 0.17 - 0.23%.

Silicon: 0.50 - 2.00%

[0032] Silicon is a compulsory element in the steel composition according to the invention to obtain the microstructure to be described. Its main function is to prevent carbon from precipitating in the form of iron carbides (most commonly cementite) and to suppress decomposition of residual austenite. Silicon contributes to the strength property and to an appropriate transformation behaviour. Additionally silicon contributes to improving the ductility, work hardenability and stretch flange formability through restraining austenite grain growth during annealing. A minimum of 0.50% Si is needed to sufficiently suppress the formation of carbides. However, a high silicon content results in formation of silicon oxides on the strip surface, which deteriorate the surface quality, coatability and workability. In addition, the Ac3 temperature of the steel composition increases as the silicon content is increased. This may affect the possibility of producing the steel strip using existing production lines in view of the maximum top temperature that can be achieved in the annealing section. In view thereof the silicon content is 2.00% or less. Preferably, Si is in the range of 0.80 -1.80% in view of wettability in combination with suppression of carbide formation and promotion of austenite stabilisation. More preferably, Si is 1.00 - 1.60%.

Aluminium: 0.01 - 0.60%

[0033] The primary function of aluminium is to deoxidise the liquid steel before casting. For deoxidation of the liquid steel 0.01% of Al or more is needed. Furthermore, aluminium has a function similar to silicon to prevent the formation of carbides and to stabilize the retained austenite. Al is deemed to be less effective compared to Si. It has no significant effect on strengthening. Small amounts of Al can be used to partially replace Si and to adjust the transformation temperatures and the critical cooling rates to obtain acicular ferrite (AF) and to accelerate the bainitic transformation kinetics. Al is added for these purposes. Therefore, the Al content is preferably more than 0.03%. High levels of Al can increase the ferrite to austenite transformation point to levels that are not compatible with current facilities, so that it is difficult to obtain a microstructure wherein the main phase is a low-temperature transformation product. The risk of cracking during casting increases as the Al content is increased. In view thereof, the upper limit is 0.60%, preferably 0.50%.

[0034] Regarding the relation between the proportions of Si and Al the composition meets the condition $Si + Al \geq 0.60$, preferably $Si + Al \geq 1.00$. Advantageously the content of Al is less than 0.5 times the Si content.

Manganese: 1.70 - 3.00%

[0035] Manganese is required to obtain the microstructure in the steel strip according to the invention in view of hardenability and stabilisation of the retained austenite. Mn also has an effect on the formation of pro-eutectoid ferrite at higher temperatures and the bainitic ferrite transformation kinetics. A certain amount of Si and/or Al is necessary to suppress the carbide formation in the bainitic ferrite. The Ac3 temperature increases as the content of Si and Al is increased. Mn is also adjusted to balance the elevated phase transformation point Ac3 as a result of the presence of Si and Al. If the Mn content is less than 1.70%, the microstructure to be described is difficult to obtain. Therefore, Mn needs to be added at 1.70% or more. However, if Mn is present in an excessive amount, macro-segregation is likely to occur, which results in unfavourable band formation in steels. Furthermore excessive amounts of Mn lead to slow bainitic transformation kinetics, which results in a too large amount of fresh martensite, and as a consequence the stretch flange formability is also deteriorated. Therefore, the Mn content is 3.00% or less, and preferably 2.80% or less, and more preferably $1.80 \leq Mn \leq 2.80\%$.

[0036] In order to obtain a strength of 980 MPa with an available production line, Mn, Cr and C are added in appropriate amounts, advantageously the following relationships are met: $10*C + Mn + Cr \geq 3.85$ and $8.5 \leq (Mn + Cr)/C \leq 16$.

Phosphor: < 0.050%

[0037] Phosphor is an impurity in steel. It segregates at the grain boundaries and decreases the workability. Its content is less than 0.050%, preferably less than 0.020%.

Sulphur: < 0.020%

[0038] Sulphur is also an impurity in the steel. S forms sulphide inclusions such as MnS that initiates cracks and deteriorates the stretch flange formability of the steel. The S content is preferably as low as possible, for example below 0.020%, preferably below 0.010% and more preferably less than 0.005%.

Nitrogen: < 0.0080%

[0039] Nitrogen is another inevitable impurity in steel. It precipitates as nitrides with micro alloying elements and is present in solid solution to contribute to strengthening. Excess nitrides deteriorate elongation, stretch flangeability and bendability. Therefore, advantageously the nitrogen content is 0.0080% or less, preferably 0.0050% or less, more preferably 0.0040% or less.

[0040] The steel composition may comprise one or more optional elements as follows:

Copper: 0 - 0.20%

[0041] Copper is not needed in embodiments of the steel composition, but may be present. In some embodiments, depending on the manufacturing process, the presence of Cu may be unavoidable. Copper below 0.05% is considered a residual element. Copper as alloying element may be added up to 0.20% to facilitate the removal of high Si scales formed in the hot rolling stage of manufacturing the starting steel strip and to improve the corrosion resistance when the cold rolled steel strip is used as such without surface treatment or in case of a Zn coated strip to improve the wettability by molten zinc. Cu can promote bainitic structures, cause solid solution hardening and precipitate out of the ferrite matrix, as ε-copper, thus contributing to precipitation hardening. Cu also reduces the amount of hydrogen penetrating into the steel and thus improve the delayed fracture characteristic. However, Cu causes hot shortness if an excess amount is added. Therefore, when Cu is added, the Cu content is less than 0.20%.

Chromium 0 1.00%; Nickel 0 - 0.50%; Molybdenum 0 - 0.50%

[0042] Chromium, nickel and molybdenum are not required elements, but may be present as residual elements in the steel composition. The allowable level of Cr, Ni or Mo as a residual element is 0.05% for each. As alloying elements they improve the hardenability of the steel and facilitate the formation of bainite ferrite and at the same time, they have similar effectiveness that is useful for stabilizing retained austenite. Therefore, Cr, Ni and Mo are effective for the microstructural control. The Cr, Ni or Mo content in the steel is preferably at least 0.05% to sufficiently obtain this effect. However, when each of them is added excessively, the effect is saturated and the bainitic transformation kinetics becomes too slow to obtain the required microstructure in the production line with a limited overageing time. Therefore, the amount of Cr is limited to a maximum of 1.00%. Ni is merely used to reduce the tendency of hot shortness when a relatively high amount of Cu is added. This effect of Ni is appreciable when the Ni content is > [Cu(%)/3]. The amount of Ni and Mo, if present, is limited to a maximum of 0.50% for each.

Niobium 0 - 0.100%; Vanadium 0 - 0.100%; Titanium 0 - 0.100%

[0043] The allowable level of niobium, vanadium and titanium as residual elements is 0.005% for each. One or more of niobium, vanadium and titanium may be added to refine the microstructure in the hot rolled intermediate product and the finished products. These elements possess a precipitation strengthening effect and may change the morphology of the bainitic ferrite. They have also a positive contribution to optimization of application depending properties like stretched edge ductility and bendability. In order to obtain these effects the lower limit for any of these elements, if present should be controlled at 0.005% or more. The effect becomes saturated when the content exceeds 0.10% for each of Nb and Ti and V. Therefore, when these elements are added, the contents thereof are controlled between 0.005% and 0.100%. Preferably, the upper limit is 0.050% or less for Nb and Ti and 0.100% of less for V, because if added excessively, carbide is precipitated too much resulting in deterioration of the workability. In addition, the sum of Ti + Nb + V preferably does not exceed 0.100% in view of workability and cost.

Boron 0 - 0.0030%

[0044] Boron is another optional element which, if added, is controlled between 0.0003% and 0.0030%. The allowable level of B as a residual element is 0.0003%. An addition of boron increases the quench hardenability and also helps to increase the tensile strength. To obtain these effects of B, a lower limit of 0.0003% is needed, preferably 0.0005%. However, when too much B is added, the effect is saturated. Advantageously B is controlled at 0.0025% or less, preferably

0.0020% or less.

[0045] In another preferred embodiment of the invention, Ti and/or Nb and/or V and/or Ni and/or Cu and/or Cr and/or Mo and/or B are not added as alloying elements in order to reduce the cost of the final product while still obtaining a cold rolled high strength steel strip having desired properties.

Calcium 0 - 0.0050%; rare earth elements (REM) 0 - 0.0100%

[0046] Furthermore, the composition according to the invention may optionally contain one or two elements selected from Ca and a rare earth metal (REM), in an amount consistent with a treatment for MnS inclusion control. If present as a residual element, the allowable level is 0.0005%. If added as an alloying element, Ca is controlled to a value less than 0.0050% and REM is controlled to a value less than 0.0100%. Ca and/or REM combines with sulfur and oxygen, thus creating oxysulfides that do not exert a detrimental effect on ductility, as in the case of elongated manganese sulfides which would form if no Ca or REM is present. This effect is saturated when Ca content is higher than 0.0050% or the REM content is higher than 0.0100%. Preferably the amount of Ca, if present, is controlled to a value below 0.0030%, more preferably below 0.0020%. Preferably the amount of REM, if present, is controlled to a value below 0.0080%, more preferably below 0.0050%.

[0047] The remainder of the steel composition comprises iron and inevitable impurities.

[0048] The chemical composition of the steels according to the invention matches the capacity of conventional continuous production lines.

Microstructure

[0049] The cold rolled steel strip that has been heat treated according to the invention has a complex microstructure comprising 5 - 30% of polygonal ferrite (PF), acicular ferrite (AF) and higher bainitic ferrite (HBF), wherein polygonal ferrite (PF) is at most 10%, and 50 -85% of lower bainitic ferrite (LBF) and partitioned martensite (PM), 5 - 20% retained austenite (RA) and fresh martensite (M) in an amount of 0 - 15%.

[0050] In this invention, the microstructures are functionally grouped in such a way that could be observed using optical microscopy and scanning electron microscopy. The polygonal ferrite (PF) refers to the ferrite formed at intercritical annealing or during slow cooling at temperatures above Bs. The acicular ferrite (AF) refers to the ferrite formed during cooling at temperatures between Bs and Ms. The high temperature bainitic ferrite (HBF) is the bainitic ferrite formed during austempering at a temperature between Bs and Bn. The low temperature bainitic ferrite (LBF) is the bainitic ferrite formed during austempering at a temperature between Bn and Ms. The partitioned martensite (PM) refers to the martensite formed during fast cooling (quenching) and overageing (partitioning) heat treatment.

Bainitic ferrite and partitioned martensite

[0051] The PM is obtained during quenching and partitioning when the quenching stop temperature is between Ms and Mf and the partition is conducted in the temperature range between the quenching stop temperature and Bn. The BF is obtained by transformation of the untransformed austenite during partitioning (overageing). The amount of PM depends on the quenching temperature. The amount of BF is a function of the partition temperature and time. Here it is noted that in this application the expression "partitioned martensite" is used instead of tempered martensite. Generally in metallurgy tempered martensite contains some carbide precipitates resulting from tempering. In the modified quenching and partitioning process according to the invention, because of the presence of Si and Al and because of a very short duration in the partition process, carbide formation is retarded during overageing. As a result, the carbon partitions from martensite to austenite which leads to carbon enriched retained austenite with higher stability and partitioned martensite is carbide free. BF is present in the form of plates with an ultrafine grain size. PM has a similar substructure to BF but with a finer size of the ferrite lath and consequently a finer size of retained austenite is obtained. The precipitation of carbides between the ferritic laths, which is known to be detrimental to ductility, is suppressed by alloying with Si and/or Al. The bainitic ferrite is carbide-free, in contrast conventional bainite contains carbides. Bainitic ferrite also differs from (pro-eutectoid) ferrite that has a low density of dislocations.

[0052] The carbide free BF and PM microstructures provide high strength due to the intermediate hard ferrite structure with a high dislocation density and a supersaturated carbon content. The bainitic ferrite structure also contributes to the desired high elongation, since it is carbide-free and the fine residual austenite grains can be present at the boundary of lath-shaped bainitic ferrite.

[0053] In the invention the bainitic ferrite is divided into two kinds thereof: bainitic ferrite formed at a high temperature range between Bs and Bn, referred to as high bainitic ferrite (HBF) and bainitic ferrite formed at a low temperature range between Bn and Ms, referred to as low bainitic ferrite (LBF). HBF has an average aspect ratio (defined as the length of the minor axis divided by the length of the major axis) higher than 0.35, and LBF has an average aspect ratio lower than 0.35

when the cross section of the steel strip subjected to 3% Nital etching is observed by a scanning electron microscopy with EBSD analysis. The reason to make this distinction is that bainitic ferrite formed at the higher temperature range above Bn (HBF) is similar to AF in grain size and shape and it is difficult to distinguish HBF from AF using SEM. Just like AF, HBF has a larger grain size, lower dislocation density and is softer than LBF and it acts to increase the elongation of the steel. On the other hand, LBF has a higher strength than that of HBF due to finer plate size, contributing to strength of the steel strip and also enhancing the formability. As PM has a similar microstructure to LBF except that the size of the ferrite lath and retained austenite is becoming smaller as the formation temperature is decreased. However, this change is gradual so that LBF and PM cannot be clearly distinguished by SEM observation. In this invention, LBF and PM are grouped as one microstructure as their contributions to the steel properties are also similar.

[0054] A feature of the high strength steel strip according to the present invention is that bainitic ferrite may have a composite microstructure including HBF and LBF+ PM. Therefore a high strength cold rolled steel strip with a high elongation and good formability can be obtained.

[0055] To obtain a good balance of high strength and formability, 50 - 85% LBF + PM is needed. If LBF + PM are present in excessively small amounts, the steel strip has insufficient strength. However, if LBF + PM are present in excessively large amounts, the effects of the other ferrites (PF, AF and HBF) and retained austenite regarding elongation may be compromised. Therefore the sum of LBF and PM is in the range of 50 - 85%, preferably 55 - 80%. The PM formed in the quenching step can accelerate the BF transformation kinetics of the untransformed austenite during overaging. To ensure the bainitic transformation can complete in the duration available in typical current production lines, the amount of PM can be regulated by controlling quenching stop temperature below the Ms point of the steel. The lower the quenching stop temperature is, the more PM is formed. For steels containing higher contents of alloying elements, a higher amount of PM is required.

[0056] The formation of the HBF in the current invention is due to the heating of the strip through the latent heat produced by bainitic transformation or due to heating by applying a hot dip galvanization process. The formation of HBF, if any, in the present invention allows to accelerate the bainitic transformation if necessary, such that the bainitic transformation can be completed in the limited time span in the overageing section in an existing production line. Depending on the amount of PF and AF resulting from the cooling stage, the amount of HBF is controlled, such that the total amount of PF, AF and HBF is 5 - 30 %, preferably 10 - 25%. As described above, HBF has a similar function to that of PF and AF. If sufficient amount of PF and AF has been formed in the cooling section, and for the purpose to obtain steel strip with a higher strength, the amount of HBF should be minimized to 0%. In the case that the amount of PF and AF is not sufficient, the amount of HBF can be increased. However, the amount of HBF should be controlled so that the total amount of PF, AF and HBF is 5 - 30%, preferably 10 - 25%.

Polygonal ferrite and Acicular ferrite

[0057] Proeutectoid ferrite is softer than bainitic ferrite and functionally increase the elongation of the steel strip. A certain amount of proeutectoid ferrite is introduced and the characteristics of the ferrite is controlled to increase the bainitic transformation kinetics and to enhance the stability of the retained austenite and to further increase the elongation. Two types of proeutectoid ferrite can be produced using the invention during cooling depending on the formation temperature. The ferrite phase formed during cooling at a high temperature above the Bs temperature in the slow cooling section is polygonal or blocky, called polygonal ferrite (PF). This type of ferrite has proven to increase the elongation but to decrease the yield strength and the flange formability such as the hole expansion capacity (HEC) in the presence of bainitic or martensitic phases. Ferrite formed at lower temperatures in the fast cooling section in a temperature between Bs and Ms has a near acicular shape and a smaller grain size than that of PF, and is referred to as acicular ferrite (AF). It is similar to HBF in morphology but has a relatively lower amount of dislocations. The presence of AF can increase the elongation without sacrificing strength and formability. As PF, AF and HBF have a similar function to tensile properties in the steel according to the invention, three types of these ferritic microstructures can be present, or one or two of them is/are present. For the purpose of ensuring high elongation, the volume fraction of the PF, AF and HBF is 5% or higher, preferably 10% or higher. However, if the content of these ferritic microstructures is too high and exceeds 30%, the HEC is significantly reduced. In any case, the total amount of PF, AF and HBF should be controlled to be less than 30%, preferably less than 25%. As PF is detrimental to formability such as hole expansion capacity and bendability, the amount of PF should be 10% or less, preferably 5% or less, more preferably 0% to obtain a steel with a good combination of the elongation and HEC value.

Residual austenite

[0058] The residual austenite (also known as retained austenite) refers to a region that shows a FCC phase (face-centred cubic lattice) in the final microstructure. Retained austenite enhances ductility partly through the TRIP effect, which manifests itself in an increase in uniform elongation. The volume fraction of residual austenite is 5% or higher, preferably

7% or higher to exhibit the TRIP effect. Below 5% the desired level of ductility and uniform elongation will not be achieved. The upper limit is mainly determined by the composition and processing parameters in a production line. For a given composition, the carbon content in the retained austenite becomes too low if the amount of the retained austenite is too high. Then the retained austenite is insufficiently stable and the local ductility (stretch flange formability) might be reduced to an unacceptable level. Therefore, the upper limit of the volume fraction of retained austenite is 20%, preferably 15%.

[0059] The concentration of carbon in the residual austenite has an impact on the TRIP characteristics. The retained austenite is effective in improving the elongation property, in particular when the carbon concentration in the retained austenite is 0.90 wt.% or higher. If the carbon content is too low, the retained austenite is not stable enough to produce the TRIP effect. Therefore, advantageously the carbon content in the retained austenite is 0.90 wt.% or higher, preferably 0.95 wt.% or higher. While the concentration of carbon in the retained austenite is preferably as high as possible, an upper limit of about 1.6% is generally imposed by practical processing conditions. The carbon content and the stability of the retained austenite can be adjusted by controlling the amount of ferrites.

Martensite

[0060] Martensite (M) is freshly formed in the final cooling section after austempering. It suppresses yield point elongation and increases the work hardening coefficient (n-value), which is desirable for achieving stable, neck-free deformation and strain uniformity in the final pressed part. Even at 1% of fresh martensite in the final steel strip a tensile response and thus press behaviour can be achieved comparable to conventional dual phase steels. However, the presence of the fresh martensite will impair formability due to the crack formation along the martensite and LBF/HBF interfaces. Therefore, the amount of the fresh martensite controlled to 15% or less, preferably 10% or less.

Carbides

[0061] Carbides can be present as fine precipitates, which are formed during austempering if the overaging temperature is too high or the overaging time is too long or in the form of pearlite formed during cooling if the cooling rate is too slow. According to the invention, the microstructure of the invented steel is pearlite-free and carbide-free. Pearlite-free means that the amount of the layered microstructure including cementite and ferrite is less than 5%. Carbide-free means that the amount carbide is below the detection limit of standard x-ray measurements.

Characterization of microstructures

[0062] The microstructural constituents classified in the invented steel as described above can be quantitively determined by techniques described hereafter. The volume fraction of the constituents is measured by equating the volume fraction to the area fraction and measuring the area fraction from a polished surface using a commercially available image-processing program or a suitable other technique.

[0063] PF, fresh M, RA and pearlite can be distinguished using optical microscopy (OM) and scanning electron microscopy (SEM). When a sample etched with 10% aqueous sodium metabisulfite (abbreviated SMB) is characterised under OM, pearlite is observed as dark areas, PF is observed as tinted grey areas and fresh martensite is observed as light brown areas. When a sample etched with 3% Nital solution is characterised with SEM, PF is observed as grains with a smoother surface that do not include the retained austenite, pearlite is observed as layered microstructure including both cementite and ferrite. The rest microstructure is observed as grey areas, featured by plate or lath like ferritic substructures, in which the RA is dispersed in the grains as white or pale grey areas and no carbides can be identified. This microstructural group is referred to as the bainitic ferrite like microstructure. It may include a mixture of HBF, LBF, AF and PM. These microstructures cannot be clearly distinguished by using OM and SEM because their morphologies are similar.

[0064] In this invention, the bainitic ferrite like microstructure is further separated into two distinct groups by means of Electron Back Scatter Diffraction (EBSD). The first group consists of PM and LBF and the second group consists of AF and HBF. From measured EBSD data, the retained austenite can be first distinguished from the other microstructures by creating Fe(γ) partition from Fe(α). The fresh martensite (M) is then separated from the bainitic ferrite like microstructure by splitting the Fe(α) into a partition with a high average image quality (IQ) and a partition with a low average IQ. The low IQ partition is classified as martensite and the high IQ partition is classified as the bainitic ferrite like microstructure. The method of distinguishing the types of two groups is described below with reference to Fig. 1. In the bainitic ferrite (high IQ partition), regions having a difference in orientation not lower than 15° in the inclination angle between adjacent structures are identified. A region is regarded as having the same crystal orientation and is defined as a bainitic plate in the present invention. For the bainitic plates thus detected, the diameter of a circle that has the same area as a bainitic plate is determined. The diameter of the equivalent circle of the bainitic plate is determined by using the photograph of EBSD analysis with magnification factor of 3000. By fitting an ellipse to a bainitic plate, the aspect ratio (defined as the length of the minor axis divided by the length of the major axis) is also determined. Similarly, diameters of the equivalent circles of all bainitic plates and aspect ratios of the equivalent ellipses of all bainitic plates in the measured area (about 100 by 100 μm) are measured and the average values are defined as the mean grain size of bainitic plates and the mean aspect ratio of the bainitic plates in the present invention.

[0065] The inventors have systematically studied the effect of the austempering temperature on the microstructure of the bainitic ferrite. The austempering temperature ranges from Ms - 200 to Bs. It has been found that the mean size and the mean aspect ratio of the bainitic plates increase as the austempering temperature is increased. Especially, the aspect ratio of the bainitic plates is found to have a sharp change between the samples austempered below 440 °C, which is below Bn and above 460 °C, which is above Bn of the steel composition used in the method according to the invention. Thus, the critical mean value of the aspect ratio of 0.35 is defined to split the two groups of bainitic ferrite like microstructure. The group consisting of LBF and PM has an aspect ratio of 0.35 or less and the group consisting of HBF and AF has an aspect ratio of more than 0.35.

[0066] In addition to the difference in the morphology and the size of the bainitic plates, the misorientation relationships among the intricate crystallographic plates between the HBF, AF group and the LBF, PM group are also different. The misorientation angle distribution in the steel according to the invention is shown in Fig. 2. The peak at 60° is consistent with the misorientations between neighbouring grains, bearing Kurdjumov-Sachs (KS/KS) relationship, which is caused by the axe-angle relationship 60°<111> and 60°<110> and corresponds to martensite. The peak at 53° - 54° is due to the misorientations between grains obtained by phase transformations according to the relationship of Nishiyama-Wassermann and Kurdjumov-Sachs (NW/KS). According to prior art (see A.-F. Gourgues, H. M. Flower, and T. C. Lindley, Materials Science and Technology, January 2000, Vol. 16, p. 26-40), acicular ferrite and upper bainite grow with Nishiyama-Wassermann relationships with the parent austenite phase, whereas lower bainite and martensite consist of highly intricate packets having Kurdjumov-Sachs relationships with the parent phase. In analogy with these results, it is assumed that the peak at 53 - 54° corresponds to the formation of HBF and AF, and the peak at 60° corresponds to the formation of LBF and PM. The peak at 53 - 54° becomes more distinguishable and the height of the peak increases but the height of the peak at 60° decreases as the austempering temperature is increased. In the present invention, the relative amounts of the HBF, AF group and the LBF, PM group can be determined by the ratio of the height of the two peaks.

[0067] As some of the retained austenite is dispersed as film in very small size between the bainitic plates and cannot be detected by EBSD, the fraction of the retained austenite determined by EBSD is always lower than the actual value. Therefore, an intensity measuring method based on XRD as a conventional technique of measuring content of retained austenite can be employed. The volume fraction of retained austenite is determined at ¼ thickness of the steel strip. The amount of cementite is also measured from this XRD analysis. A sample prepared from the steel strip is mechanically and chemically polished and is then analyzed by measuring the integral intensity of each of the (200) plane, (220) plane, and (311) plane of fcc iron and that of the (200) plane, (211) plane, and (220) plane of bcc iron with an X-ray diffractometer using Co-Ka. The amount of retained austenite (RA) and the lattice parameter in the retained austenite were determined using Rietveld analysis. The C content in the retained austenite is calculated using the formula:

$$C\ (\text{wt.\%}) = (a[\text{Å}] - 3.572 - 0.0012\ \text{Mn\%} + 0.00157\ \text{Si\%} - 0.0056\ \text{Al\%})/0.033$$

where a is the lattice parameter of the retained austenite in angstrom.

Mechanical properties

[0068] The cold rolled steel strips with the above microstructure and composition and heat treated according to the invention have such properties:

Yield strength (YS) is at least 550 MPa; and/or
Tensile strength (TS) is at least 980 MPa; and/or
Total elongation (TE) is at least 13%; and/or
Hole expansion capacity (HEC) is at least 20%; and/or
Bending angle (BA) is at least 80°.

[0069] Preferably the cold rolled and heat treated strip possesses all these properties.

Method steps

[0070] According to the method of the invention a cold rolled steel strip having the composition as explained above is heat treated to obtain the microstructure and properties. The cold rolled steel strip obtained through cold rolling is subjected to a thermal treatment as in a continuous annealing line. A typical design of the process is diagrammatically shown in Fig. 3. The cold rolled steel strip is heated above the temperature (Ac3 - 20), e.g. using a heating rate of at least 0.5 °C/s, preferably to the temperature range of (Ac3 - 20) - (Ac3 + 20), typically to a predetermined austenization temperature T2, and held for a period of time t2 within this temperature range (step a), and then cooled, typically using a two-step cooling at controlled cooling rates, to a temperature T4 below Ms, typically in the range of Ms - (Ms - 200) (step b). Then the steel strip is heated (step c), which optionally involves a heat treatment below Ms, typically in the range T4 - Ms, to above Ms and subsequently treated in the range of Ms - Bs for austempering for a time t5 (step d), typically at a temperature T5 in the range of Ms to Bn. Optionally, the steel strip is then heated to a temperature T6 in the range of Bn to Bs for a period of time t6, which may be a temperature at which a hot dip galvanizing treatment is possible. Finally, the steel strip is cooled down to room temperature (step e). The process parameters and functions in each step will be described hereinafter.

[0071] In a first step the cold rolled steel is soaked above (Ac3 - 20), such as within a temperature range of (Ac3 - 20) - (Ac3 + 20) °C, during a soaking time t2 of 1 - 200 seconds in order to achieve a fully austenitic microstructure. Annealing at a temperature above (Ac3 - 20) is necessary because the steel strip that is heat treated according to the invention, needs to have the required amounts of the low temperature transformed phases such as bainitic ferrite and retained austenite, as well as a predetermined amount of ferrite, which are transformed from high temperature single austenite phase. If T2 is lower than (Ac3 - 20) or the annealing time t2 is shorter than 1s, reverse transformation to austenite may not proceed sufficiently and/or carbides in the steel sheet may not be dissolved sufficiently and a single austenite phase microstructure is not ensured. If T2 is higher than (Ac3 + 20) or t2 is longer than 200 seconds, austenite grains will grow, which influences the size and distribution of the retained austenite and also slows down the bainitic transformation kinetics later in the overaging process. An excess amount of fresh martensite formed during final cooling may form as a result of this incomplete bainitic transformation, which leads to a higher strength but a low ductility and formability. Moreover,, a uniform single austenite structure with larger grain sizes may suppress the formation of PF and AF in the following cooling section so that an insufficient amount of ferrite is obtained within the current cooling schedule in the available production line, and may cause the steel strip to have an insufficient elongation. It has been observed that the uniformity of the austenite has a large effect on the formation of PF and AF in the cooling section. Accordingly, the annealing temperature needs to be higher than (Ac3 - 20), but advantageously not to exceed (Ac3 + 20), preferably in the range of (Ac3 - 15) to (Ac3 + 15). The annealing time t2 is 1 second to 200 seconds, preferably 40 seconds to 150 seconds.

[0072] In a subsequent cooling step the austenitic strip is cooled to a temperature T4 below Ms, typically in the range of Ms to Ms - 200. The purpose of this cooling is to regulate the amounts of ferrites and partitioned martensite, but to prevent the formation of pearlite.

[0073] The cooling is realized by a two-step cooling in order to regulate the amount of ferrite and to homogenize the strip temperature. This fits most of the continuous annealing lines or hot dip galvanizing lines which include two connected cooling sections as currently in use. The steel strip is first cooled to a temperature T3 in the range of 800 - 550 °C (referred to as slow cooling section), preferably in the range of 750 - 550 °C, at a cooling rate of V3 of 2 - 15 °C/s, preferably 3 - 10 °C/s. Thereafter, the steel strip is cooled further down to the temperature T4 (referred to as fast cooling section), at a cooling rate V4 of 20 - 70 °C/s. As the length at each section in a continuous annealing line is fixed, the cooling rates V3 and V4 for a given line speed can be controlled by adjusting the T3 temperature. The higher the T3 is, the lower the V3 is and the higher the V4 is. During this cooling some PF may be formed in the slow cooling section, and some AF may be formed in the fast cooling section. For a fixed line speed, the amount of PF formed in the slow cooling section mainly depends on T3 and the amount of AF mainly depends on V4.. Therefore T3 is selected in a suitable range to adjust the amount of ferrite and to prevent the formation of pearlite. If T3 is too low, e.g. lower than 550 °C,PF may form in an excess amount in the slow cooling section and AF may also form in an excess amount in the fast cooling section, or even pearlite may form if the resulting V4 is lower than 15 °C/s. If T3 is too high, e.g. higher than 800 °C, PF may form insufficiently and less AF is formed if the resulting V4 is too high. Accordingly, T3 should be in the range of 800 to 550°C, preferably in the range of 750 to 600 °C.

[0074] After cooling to the temperature T4 below Ms, preferably in the range of Ms - (Ms - 200), some amount of martensite is obtained. The lower T4 is, the more martensite is formed. To effectively accelerate the bainitic transformation kinetics in the following partition process, T4 is adjusted according to the steel compositions. For steels containing higher amounts of alloying elements, a lower T4 is applied. If T4 is too high, an insufficient amount of PM is formed. The bainitic transformation of the untransformed austenite could not be completed in the overageing (partitioning) stage and too much fresh martensite may form in the following cooling process to ambient temperature. If T4 is too low, too much PM is formed and the amount of the retained austenite is reduced. Therefore, T4 is preferably in the range of Ms - (Ms - 200), more preferably (Ms - 50) - (Ms - 150). As the amount of PM only depends on the T4 temperature, the steel strip is heated as fast as possible to the partition temperature in the range of Ms - Bs in order to allow utilization of the remainder of the totally available time span in the overageing section for the bainitic transformation. In practice, depending on the heating capacity of a production line and to facilitate homogenisation of the temperature of the steel strip, the total duration t4 of step c) including any optional holding time is preferably less than 10s, more preferably less than 5s. Optionally heating step c) may involve a brief heat treatment in the temperature range below Ms, for example in the range of Ms - (Ms - 200), such as in the temperature range of (Ms - 50) - (Ms - 150).

[0075] In the subsequent heat treatment step d) the cooled strip is heat treated at a temperature T5 above Ms and below Bs, preferably below Bn for a time t5 in the range of 30 - 120 seconds. By heating to and heat treating at a temperature T5 in this range, the untransformed austenite transforms into lower bainitic ferrite (LBF) and carbon partitioning occurs in the prior formed martensite. If T5 is too low, the bainitic transformation is too slow, the bainitic transformation is insufficient during overageing and fresh martensite may form during cooling after overageing in excessive amounts, which increases the strength but reduces the required elongation. On the other hand, carbon partitioning may be insufficient to stabilize the retained austenite. If T5 is too high too much HBF is obtained in the overageing section, which cannot provide the required strength. The preferred range for T5 is (Bn - 50) to Bn in order to achieve the fast bainitic transformation kinetics. If the heat treatment time t5 is less than 30s, the bainitic transformation is incomplete and also the carbon partitioning in martensite and bainite is insufficient. If t5 is more than 120s, there is a risk that carbides start to form and therefore decrease the carbon content in the retained austenite. The maximum time for t5 is limited by inter alia the total available time at a given speed of the production line. Preferably, t5 is in the range of 40 to 100 seconds.

[0076] As the steel strip temperature can be increased by latent heat produced by bainite transformation during overageing, . a small amount of high temperature bainitic ferrite will be formed if the steel strip reaches temperatures higher than Bn.

[0077] Subsequently the thus heat treated strip is cooled following the production line capacity to ambient temperature during which some fresh martensite may be formed. The steel strip is then cooled down to below 300 °C at a cooling rate V7 of at least 1 °C/s, preferably at least 5 °C/s, after which it is further cooled down to ambient temperature. Cooling down to ambient temperature may be forced cooling or uncontrolled natural cooling. In a practical embodiment the heat treated steel strip is cooled to a temperature T7 in the range of (Ms - 50) - Mf at a cooling rate V7 in the range of 5.0 - 10.0 °C/s. Further cooling from T7 to ambient temperature is preferably performed at a cooling rate V8 of 5.0 - 20.0 °C/s, more preferably 6.0 - 15.0 °C/s.

[0078] Advantageously the heating step, prior to the soaking step, is performed in two substeps, comprising heating a cold rolled strip to a temperature T1 in the range of 680 - 740 °C, preferably in the range of 700 - 720 °C, at a heating rate V1 of 10.0 - 30.0 °C/s, preferably 15.0 - 25.0 °C/s; and further heating the cold rolled strip from the temperature T1 to the soaking temperature T2 at a heating rate V2 of 0.5 - 4.0 °C/s,, preferably 1.0 - 3.0 °C/s. During the slow heating from T1 to the soaking temperature T2, recovery and recrystallization occur in the ferrite, as well as dissolution of carbides and ferrite during austenite transformation. T1 and V2 affect the progress of these processes, which affect the austenite grain size and the homogeneity of the distribution of the alloying elements in the austenite phase. Advantageously the soaking time t2 is controlled, depending on the heating rate V2, to ensure dissolution of all carbides and avoidance of a coarse austenitic grain size.

[0079] In an embodiment the method according to the invention comprises a further heat treatment step between the heat treatment step d) and cooling step e), wherein the steel strip resulting from step d) is subjected to an additional heat treatment in the range of Bs - Bn, preferably (Bs - 50) - Bn, typically at a fixed temperature T6. The additional treatment time t6 is advantageously 5 - 30 seconds, preferably 10 - 20 seconds. This additional heat treatment increases the bainitic ferrite by formation of high temperature bainitic ferrite from remaining austenite to complete the bainitic transformation and therefore further reduces the amount of martensite formed in the following cooling section, enabling improvement of the strength and ductility properties. Carbon also further partitions into the retained austenite making it more stable. When this additional heat treatment is applied in a given overageing section and thus a given total time span therein, the time t5 is further reduced to meet the available time span, e.g. the sum of t4 + t5 + t6 is in the range of 30 - 120s.

[0080] In a preferred embodiment this additional heat treatment comprises an integrated hot dip galvanizing treatment, wherein the steel strip resulting from step c) is coated with a Zn or Zn alloy based coating.

[0081] The steel strip that has been heat treated according to the invention can be provided with a coating, advantageously a zinc or zinc alloy based coating. Advantageously the zinc based coating is a galvanized or galvannealed coating. The Zn based coating may comprise a Zn alloy containing Al as an alloying element. A preferred zinc bath composition contains 0.10-0.35% Al, the remainder being zinc and unavoidable impurities. Another preferred Zn bath comprising Mg and Al as main alloying elements, has the composition: 0.5 - 3.8% Al, 0.5 - 3.0% Mg, optionally at most 0.2% of one or more additional elements; the balance being zinc and unavoidable impurities. Examples of the additional elements include Pb, Sb, Ti, Ca, Mn, Sn, La, Ce, Cr, Ni, Zr and Bi.

[0082] The coating such as a protective coating of Zn or Zn alloy may be applied in a separate step. Preferably a hot dip galvanizing step is integrated in the method according to the invention as explained above.

[0083] Optionally a temper rolling treatment may be performed with the annealed and zinc coated strip according to the invention in order to fine tune the tensile properties and to modify the surface appearance and roughness depending on the specific requirements resulting from the intended use.

[0084] The cold rolled steel strip as such is typically manufactured according to the following general process. A composition as described above is prepared and cast into a slab. The cast slab is processed using hot rolling after reheating at a temperature in the range of 1100 - 1300 °C. Typically hot rolling of the slab is performed in 5 to 7 stands to final dimensions that are suitable for further cold rolling. Typically finish rolling is performed in the fully austenitic condition above 800 °C, advantageously 850 °C or higher. The strip thus obtained from the hot rolling steps may be coiled, e.g. at a coiling temperature of typically 700 °C or lower. The hot rolled strip is pickled and cold rolled to obtain a cold rolled steel strip with proper gauges. Preferably the cold rolling reduction is in the range of typically 30 to 80%. In order to reduce the rolling force during cold rolling, the coiled strip or half cold rolled strip may be subjected to hot batch annealing. The batch annealing temperature should be in the range of 500 - 700 °C.

[0085] Thin slab casting, strip casting or the like can also be applied. In this case it is acceptable for the manufacturing method to skip at least a part of the hot rolling process.

Examples

[0086] Steels having compositions as shown in Table 1 were cast into 25 kg ingots of 200 mm × 110 mm × 110 mm in dimensions using vacuum induction. The following process schedule was used to manufacture cold rolled strips of 1 mm thickness:

- Reheating of the ingots at 1225 °C for 2 hours;
- Rough rolling of the ingots from 140 mm to 35 mm;
- Reheating of the rough-rolled ingots at 1200 °C for 30 min;
- Hot rolling from 35 mm to 4 mm in 6 passes;
- Run-out-table cooling: Cool from finish rolling temperature (FRT) about 850 to 900 °C to 600 °C at a rate of 40 °C/s;
- Furnace cooling: Strips transferred to a preheated furnace at 600 °C and then cooled to room temperature to simulate the coiling process;
- Pickling: The hot rolled strips were then pickled in HCl at 85 °C to remove the oxide layers;
- Cold rolling: The hot rolled strips were cold rolled to 1 mm strips;
- Heat treating according to the invention: Cold rolled sheets with suitable size were used to simulate the annealing process by using a continuous annealing simulator (CASIM);Samples for microstructure observations, tensile tests and hole expansion tests were machined from the thus treated strips.

[0087] Dilatometry was done on the cold rolled samples of 10 mm × 5 mm × 1 mm dimensions (length along the rolling direction). Dilatation tests were conducted on a Bahr dilatometer type DIL 805. All measurements were carried out in accordance with SEP 1680. The critical phase transformation points Ac3, Ms and Mf were determined from the quenched dilatometry curves. Bs and Bn were predicted using available software JmatPro 10. The phase fractions during annealing for different process parameters were determined from dilatation curves simulating the annealing cycles.

[0088] The microstructure was determined by optical microscopy (OM) and scanning electron microscopy (SEM) using a commercially available image-processing program. The microstructures were observed at ¼ thickness in the cross section of rolling and normal directions of a steel strip. The Scanning Electron Microscope (SEM) used for the EBSD measurements is a Zeiss Ultra 55 machine equipped with a Field Emission Gun (FEG-SEM) and an EDAX PEGASUS XM 4 HIKARI EBSD system. The EBSD scans were captured using the TexSEM Laboratories (TSL) software OIM (Orientation Imaging Microscopy) Data Collection. The EBSD scans were evaluated with TSL OIM Analysis software. The EBSD scan area was in all cases 100 × 100 μm, with a step size of 0.1 μm, and a scan rate of approximately 80 frames per second.

[0089] The retained austenite was determined by XRD according to DIN EN 13925 on a D8 Discover GADDS (Bruker AXS) with Co-Kα radiation. Quantitative determination of phase proportions was performed by Rietveld analysis.

[0090] Tensile tests - JIS5 test pieces (gauge length = 50 mm; width = 25 mm) were machined from the annealed strips such that the tensile direction was parallel to the rolling direction. Room temperature tensile tests were performed in a Schenk TREBEL testing machine following NEN-EN10002-1:2001 standard to determine tensile properties (yield strength YS (MPa), ultimate tensile strength UTS (MPa), total elongation TE (%)). For each condition, three tensile tests were performed and the average values of mechanical properties are reported.

[0091] Hole Expansion Test (Stretch Flangeability Evaluation Test) - Test pieces for testing hole expandability (size: 90×90 mm) were sampled from the obtained rolled strip. In accordance with The Japan Iron and Steel Federation Standards JFS T 1001, a 10 mm diameter punch hole was punched in the centre of the test piece and a 60° conical punch was pushed up and inserted into the hole. When a crack penetrated the strip thickness, the hole diameter d (mm) was measured. The hole expansion ratio λ (%) was calculated by the following equation: $\lambda\ (\%)=\{(d-d0)/d0\}\times 100$, with d0 being 10 mm.

[0092] Bending test - Bending specimens (40 mm × 30 mm) from parallel and transverse to rolling directions were prepared from each of the conditions and tested by three-point bending test according to the VDA 238-100 standard. The experiments were stopped at different bending angles and the bent surface of the specimen was inspected for identification of failure in order to determine the bending angle (BA). The bending angles of the samples with bending axis parallel to the rolling direction are lower than those of the samples with bending axis perpendicular to the rolling direction. For each type of tests, three samples were tested and the average values from three tests are presented for each condition.

[0093] The process parameters are presented in Table 2 using the indications in Figure 3. In CASIM, the steel strip is cooled down at V4 to T4 and then heated to T5 within 5s.

[0094] The resulting microstructures and tensile properties are given in Table 3. Steel A53 (example 6 and 7) could not reach the required tensile strength or elongation because the condition of (10C + Mn + Cr) is not met. Examples 14 and 20 show that if the amount of the (PF + AF + HBF) is not high enough, the required elongation cannot be reached.

Table 1. Composition (in wt%) and the critical phase transformation points (in °C) of the steels

Alloy code	C	Mn	Si	Al	Cr	Cu	Nb	Mo	S	P	Ti	V	N	Ac3	Ms	Mf	Bs	Bn	10C+Mn+Cr	Note
A16	0.210	2.210	1.170	0.016	0.010	0.005	0.002	0.002	0.002	0.003	0.002	0.002	0.004	855	372	178	521	449	4.32	inv.
A50	0.207	2.290	1.005	0.037	0.010	0.005	0.001	0.002	0.003	0.002	0.002	0.002	0.002	846	385	178	525	452	4.37	inv.
A52	0.158	2.509	1.019	0.038	0.010	0.003	0.001	0.003	0.005	0.002	0.001	0.002	0.002	857	401	190	524	457	4.10	inv.
A53	0.176	2.013	1.003	0.040	0.010	0.005	0.001	0.002	0.005	0.003	0.001	0.002	0.003	862	410	198	539	460	3.78	comp.
A54	0.206	2.279	1.468	0.038	0.010	0.005	0.001	0.002	0.006	0.004	0.001	0.002	0.005	871	381	175	511	444	4.35	inv.
A72	0.207	1.787	1.446	0.039	0.010	0.001	0.001	0.002	0.003	0.003	0.001	0.001	0.003	879	408	205	529	445	3.87	inv.
A73	0.207	2.470	1.525	0.036	0.010	0.001	0.001	0.003	0.005	0.001	0.001	0.002	0.005	865	365	162	503	434	4.55	inv.
A79	0.213	1.940	1.470	0.020	0.500	0.002	0.001	0.003	0.005	0.003	0.002	0.003	0.002	863	390	195	513	446	4.57	inv.
A95	0.205	2.442	1.183	0.321	0.010	0.003	0.001	0.004	0.002	0.005	0.001	0.002	0.002	880	391	175	513	443	4.50	inv.
A99	0.250	2.299	1.021	0.036	0.010	0.003	0.001	0.003	0.004	0.003	0.001	0.003	0.002	833	358	155	518	442	4.81	inv.

Table 2. Process parameters.

Example	Alloy code	Process parameters indicated in Figure 3															
		T1	V2	T2	t2	V3	T3	V4	T4	t4	T5	t5	T6	t6	V7	T7	V8
		°C	°C/s	°C	s	°C/s	°C	s	°C	s	°C	s	°C	s	°C/s	°C	°C/s
1	A16	720	1.44	860	65	3.9	700	52.3	360	1	455	53	455	19	6.3	300	11.3
2	A50	720	1.34	850	65	3.2	720	60	330	3	440	50	460	19	6.5	300	11.3
3	A50	720	1.34	850	65	3.2	720	57	350	3	440	50	460	19	6.5	300	11.3
4	A52	720	1.55	870	65	4.2	680	60	300	5	440	48	460	19	6.5	300	11.3
5	A52	720	1.55	870	65	7.9	600	42	300	5	440	48	460	19	6.5	300	11.3
6	A53	720	1.55	870	65	4.2	700	62	300	5	440	48	460	19	6.5	300	11.3
7	A53	720	1.55	870	65	7.9	550	38	300	5	440	48	460	19	6.5	300	11.3
8	A54	720	1.65	880	65	4	720	60	330	3	420	50	460	19	6.5	300	11.3
9	A54	720	1.65	880	65	4	720	60	330	3	440	50	460	19	6.5	300	11.3
10	A54	720	1.65	880	65	4	720	57	350	3	440	50	460	19	6.5	300	11.3
11	A54	720	1.65	880	65	5	680	54	330	3	420	50	460	19	6.5	300	11.3
12	A72	720	1.7	880	65	4.4	700	61.5	300	5	440	48	460	19	6.5	300	11.3
13	A73	720	1	880	1	8.1	550	39	300	5	440	48	460	19	6.5	300	11.3
14	A79	720	1.4	860	65	3.4	720	65	300	5	400	48	460	19	6.5	300	11.3
15	A79	720	1.2	840	65	3	720	65	300	5	400	48	460	19	6.5	300	11.3
16	A79	720	1.2	840	65	3	720	65	300	5	420	48	460	19	6.5	300	11.3
17	A79	720	1.2	840	65	3	720	65	300	5	440	48	460	19	6.5	300	11.3
18	A95	720	1.55	870	65	7.6	560	40	300	5	440	48	460	19	6.5	300	11.3
19	A95	720	1.55	870	65	4.2	700	61.5	300	5	440	48	460	19	6.5	300	11.3
20	A99	720	1.24	840	65	3.5	700	61.5	300	5	440	48	460	19	6.5	300	11.3
21	A99	720	1.24	840	65	5.7	610	40	350	5	440	48	460	19	6.5	300	11.3

Table 3. Microstructures and properties

Example	Alloy code	Microstructures						Tensile properties				Formability			
		PF	PF+AF+HBF	LBF	M	RA%	C in RA	YS	TS	UE	TE	HEC	BA⊥ RD	BA//RD	Note
		%	%	%	%	%	wt.%	MPa	MPa	%	%	%			
1	A16	0	23	57	9	11	1.02	736	1173	9.6	15	22	102	98	inv.
2	A50	0	5	80.9	1	13.1	0.99	663	1064	11.7	15.6	30	97	93	inv.
3	A50	0	6	81.1	1	11.9	0.96	689	1084	10.6	15.1	34	104	91	inv.
4	A52	0	8	84.1	1	7.9	0.95	894	1009	10.41	13.9	56	110	105	inv.
5	A52	6	23	69.7	0	8.3	1	723	998	12.5	17.2	34	107	103	inv.
6	A53	0	7	89.6	0	3.4	1.09	952	1067	6.8	9.5	74	111	108	comp.
7	A53	73	75	12.8	1	11.2	1.03	373	852	15.1	21.3	15	78	75	comp.
8	A54	0	5	79.5	1	14.5	1.05	816	1135	10.2	14.6	48	102	98	inv.
9	A54	0	5	78.6	1	15.4	1.02	742	1129	10	14.3	45	105	101	inv.
10	A54	0	6	77		16.6	1.01	703	1158	9.7	13.7	42	107	99	inv.
11	A54	3	8	76	1	14.9	1.02	874	1154	10.9	16.6	42	110	102	inv.
12	A72	6	18	71	1	10	0.98	907	1093	8.9	13.3	56	98	95	inv.
13	A73	10	23	63	0	14.2	1	712	1175	12.2	16.7	27	91	83	inv.
14	A79	0	1.5	88	0	10.5	1.05	1030	1202	8.7	12.6	39	95	88	comp.
15	A79	10	25	62	2	11	1.07	976	1191	9.6	14.2	25	99	85	
16	A79	10	24	63.3	2	10.7	0.96	719	1197	11.2	14.2	28	101	92	
17	A79	10	23	60	2	15.1	0.94	862	1207	11	14.5	30	103	95	
18	A95	9	24	60.3	1	14.7	1.06	850	1109	11.6	16.1	38	113	103	inv.
19	A95	4	7	80	1	11.9	1.07	947	1157	10.1	13.9	43	101	95	inv.
20	A99	0	2	88	1	9	1.06	920	1160	8.6	11.4	63	103	96	comp.
21	A99	10	24	50.3	11	14.7	1.01	643	1193	12.5	16.8	26	107	62	inv.

BA⊥RD: bending axis perpendicular to rolling direction
BA//RD: bending axis parallel to rolling direction

- Not according to claim 1

Claims

1. A method of heat treating a cold rolled steel strip, which method comprises the steps of:

a) soaking a cold rolled steel strip above (Ac3 - 20) for a soaking time t2 of 1 - 200 seconds, thereby obtaining a cold rolled steel strip having an austenitic microstructure;
b) cooling of the soaked steel strip resulting from step a) to a temperature T4 in the range of Ms - (Ms - 200), comprising a substep of cooling the soaked steel strip resulting from step a) to a temperature T3 in the range of 800 - 550 °C at a cooling rate V3 of 2.0 - 15.0 °C/s, and a substep of cooling the soaked steel strip from the temperature T3 to T4 at a cooling rate V4 of 20.0 - 70.0 °C/s;
c) heating the cooled steel strip resulting from step b) to a temperature range of Bs - Ms;
d) heat treating the heated steel strip in the temperature range of Bs - Ms for a period of time t5 of 30 - 120 seconds;
e) cooling the heat treated steel strip to ambient temperature;
such that the steel strip has a microstructure in vol. %) comprising

polygonal ferrite (PF):	0 - 10;
polygonal ferrite (PF) + acicular ferrite (AF) + higher bainitic ferrite (HBF):	5 - 30;
lower bainitic ferrite (LBF) + partitioned martensite (PM):	50 - 85,
retained austenite (RA):	5 - 20;
martensite (M):	0 - 15;

wherein the steel strip has a composition in mass percent) comprising

C:	0.15 - 0.28;
Mn:	1.70 - 3.00;
Si:	0.50 - 2.00;
Al:	0.01 - 0.60;
P:	less than 0.050;
S:	less than 0.020;
N:	less than 0.0080;

wherein the sum (Si + Al) is ≥ 0.60; and
wherein 10C + Mn + Cr ≥ 3.85 and 8.5 ≤ (Mn + Cr)/C ≤ 16; optionally one or more elements selected from

$$0 < Cr \leq 1.00;$$

$$0 < Cu \leq 0.20;$$

$$0 < Ni \leq 0.50;$$

$$0 < Mo \leq 0.50;$$

$$0 < Nb \leq 0.10;$$

$$0 < V \leq 0.10;$$

$$0 < Ti \leq 0.10;$$

$$0 < B \leq 0.0030;$$

$$0 < Ca \leq 0.0050;$$

$$0 < REM \leq 0.0100,$$

wherein REM is one or more rare earth metals;
and the remainder being iron and inevitable impurities.

2. The method according to claim 1, wherein step c) involves heat treating the cooled strip from step b) at a temperature T4 in the temperature range of Ms - (Ms - 200), more preferably in the temperature range of (Ms - 50) - (Ms - 150), wherein preferably the total duration t4 of step c) is in the range of 1 - 10 seconds, more preferably in the range of 1 - 5 seconds.

3. The method according to claim 1 or claim 2, wherein step a) comprises soaking a cold rolled steel strip within a temperature range of (Ac3 - 20) - (Ac3 + 20), preferably within a temperature range of (Ac3 - 15) - (Ac3 + 15), preferably for a soaking time t2 of 30 - 150 s.

4. The method according any one of the preceding claims, wherein step b) comprises cooling the soaked steel strip from step a) to the temperature T4 at a cooling rate sufficient to avoid pearlite formation.

5. The method according to any one of the preceding claims, wherein step b) comprises a substep of cooling the soaked steel strip resulting from step a) to a temperature T3 in the range of 750 - 600 °C, preferably at a cooling rate V3 of 3.0 - 10.0 °C/s.

6. The method according to any one of the preceding claims, prior to step a) further comprising heating a cold rolled strip to a temperature above (Ac3 - 20) at a heating rate of at least 0.5 °C/s, preferably comprising heating the cold rolled strip to a temperature T1 in the range of 800 - 550 °C, preferably in the range of 750 - 600 °C, at a heating rate V1 of 10.0 - 30.0 °C/s, preferably at a heating rate V1 of 15.0 - 25.0 °C/s; and further heating the cold rolled strip from the temperature T1 to a temperature above (Ac3 - 20), preferably to the temperature range of (Ac3 - 20) - (Ac3 + 20), more preferably (Ac3 - 15) - (Ac3 + 15), at a heating rate V2 of 0.5 - 4.0 °C/s, preferably 1.0 - 3.0 °C/s.

7. The method according to any one of the preceding claims, wherein in step d) heat treating is performed in the range of Bn - (Ms + 50), preferably during a period of time t5 of 40 - 100 seconds.

8. The method according to any one of the preceding claims, comprising a further heat treatment step between steps d) and e) of heat treating the steel strip resulting from step c) in the range of Bs - Bn, preferably (Bs - 50) - Bn, preferably for a period of time t6 of 5 - 30 seconds, more preferably for a period of time t6 of 10 - 20 seconds.

9. The method according to claim 8, wherein the further heat treatment step comprises a hot dip galvanizing treatment.

10. The method according to any one of the preceding claims 1-8, following heat treatment further comprising a coating step of coating the heat treated steel strip with a protective coating, preferably a Zn or Zn alloy coating.

11. The method according to any one of the preceding claims, wherein the microstructure comprises in vol.%:

polygonal ferrite (PF):	0 - 5;
polygonal ferrite (PF) + acicular ferrite (AF) + higher bainitic ferrite (HBF):	10 - 25;
lower bainitic ferrite (LBF) + partitioned martensite (PM):	55 - 80;
retained austenite (RA):	7 - 15;
martensite (M):	0 - 10;

and/or wherein the C content in retained austenite (RA) is 0.90 wt. % or more, preferably 0.95 wt. % or more.

12. The method according to any one of the preceding claims, wherein the resulting steel strip has at least one, preferably all, of the properties:

Yield strength (YS) is at least 550 MPa; and/or
Tensile strength (TS) is at least 980 MPa; and/or
Total elongation (TE) is at least 13%, and/or
Hole expansion capacity (HEC) is at least 20%; and/or
Bending angle (BA) is at least 80°.

Patentansprüche

1. Verfahren des Wärmebehandelns eines kaltgewalzten Stahlbandes, wobei das Verfahren folgende Schritte umfasst:

a) Wärmeausgleichen eines kaltgewalzten Stahlbandes über (Ac3 - 20) über eine Wärmeausgleichszeit t2 von 1 - 200 Sekunden, wodurch ein kaltgewalztes Stahlband mit einer austenitischen Mikrostruktur erhalten wird;
b) Abkühlen des aus Schritt a) resultierenden ausgeglichenen Stahlbandes auf eine Temperatur T4 im Bereich von Ms - (MS - 200), umfassend einen Teilschritt des Abkühlens des aus Schritt a) resultierenden ausgeglichenen Stahlbandes auf eine Temperatur T3 im Bereich von 800 - 550 °C bei einer Abkühlrate V3 von 2,0 - 15,0 °C/s und einen Teilschritt des Abkühlens des ausgeglichenen Stahlbandes von der Temperatur T3 auf T4 bei einer Abkühlrate V4 von 20,0 - 70,0 °C/s;
c) Erwärmen des abgekühlten Stahlbandes aus Schritt b) auf einen Temperaturbereich von Bs - Ms;
d) Wärmebehandeln des erwärmten Stahlbandes in einem Temperaturbereich von Bs - Ms über eine Zeitspanne t5 von 30 - 120 Sekunden;
e) Abkühlen des wärmebehandelten Stahlbandes auf Umgebungstemperatur;
so dass das Stahlband eine Mikrostruktur (in Vol.-%) aufweist, umfassend

polygonalen Ferrit (PF):	0 - 10;
polygonalen Ferrit (PF) + nadelförmigen Ferrit (AF) + oberen bainitischen Ferrit (HBF):	5 - 30;
unteren bainitischen Ferrit (LBF) + partitionierten Martensit (PM):	50 - 85;
Restaustenit (RA):	5 - 20;
Martensit (M):	0 - 15;

wobei das Stahlband eine Zusammensetzung (in Massenprozent) aufweist, umfassend

C	0,15 - 0,28;
Mn	1,70 - 3,00;
Si	0,50 - 2,00;
Al	0,01 - 0,60;
P	weniger als 0,050;
S	weniger als 0,020;
N	weniger als 0,0080;

wobei die Summe (Si + Al) $\geq$ 0,60 ist, und
wobei 10C + Mn + Cr $\geq$ 3,85 und 8,5 $\leq$ (Mn + Cr)/C $\leq$ 16 ist;
gegebenenfalls ein oder mehrere Elemente, ausgewählt aus

$$0 < \mathrm{Cr} \leq 1{,}00;$$

$$0 < \mathrm{Cu} \leq 0{,}20;$$

$$0 < Ni \leq 0{,}50;$$

$$0 < Mo \leq 0{,}50;$$

$$0 < Nb \leq 0{,}10;$$

$$0 < V \leq 0{,}10;$$

$$0 < Ti \leq 0{,}10;$$

$$0 < B \leq 0{,}0030;$$

$$0 < Ca \leq 0{,}0050;$$

$$0 < REM \leq 0{,}0100,$$

wobei REM ein oder mehrere Seltenerdmetalle ist;
und der Rest Eisen und unvermeidliche Verunreinigungen sind.

2. Verfahren nach Anspruch 1, wobei Schritt c) die Wärmebehandlung des gekühlten Bandes aus Schritt b) bei einer Temperatur T4 im Temperaturbereich von Ms - (Ms - 200), vorzugsweise im Temperaturbereich von (Ms - 50) - (Ms - 150), umfasst, wobei vorzugsweise die Gesamtdauer t4 von Schritt c) im Bereich von 1 - 10 Sekunden, stärker bevorzugt im Bereich von 1 - 5 Sekunden liegt.

3. Verfahren nach Anspruch 1 oder Anspruch 2, wobei Schritt a) das Wärmeausgleichen eines kaltgewalzten Stahlbandes in einem Temperaturbereich von (Ac3 - 20) - (Ac3 + 20), vorzugsweise in einem Temperaturbereich von (Ac3 - 15) - (Ac3 + 15), vorzugsweise für eine Wärmeausgleichszeit t2 von 30 - 150 s, umfasst.

4. Verfahren nach einem der vorhergehenden Ansprüche, wobei Schritt b) das Abkühlen des ausgeglichenen Stahlbandes aus Schritt a) auf die Temperatur T4 mit einer Abkühlgeschwindigkeit umfasst, die ausreicht, um Perlitbildung zu vermeiden.

5. Verfahren nach einem der vorhergehenden Ansprüche, wobei Schritt b) einen Teilschritt des Abkühlens des aus Schritt a) resultierenden ausgeglichenen Stahlbandes auf eine Temperatur T3 im Bereich von 750 - 600 °C, vorzugsweise bei einer Abkühlgeschwindigkeit V3 von 3,0 - 10,0 °C/s, umfasst.

6. Verfahren nach einem der vorhergehenden Ansprüche, das vor Schritt a) ferner das Erwärmen eines kaltgewalzten Bandes auf eine Temperatur über (Ac3 - 20) mit einer Erwärmungsgeschwindigkeit von mindestens 0,5 °C/s, vorzugsweise das Erwärmen des kaltgewalzten Bandes auf eine Temperatur T1 im Bereich von 800 - 550 °C, vorzugsweise im Bereich von 750 - 600 °C, mit einer Erwärmungsgeschwindigkeit V1 von 10,0 - 30,0 °C/s, vorzugsweise mit einer Erwärmungsgeschwindigkeit V1 von 15,0 - 25,0 °C/s, umfasst, und ferner Erwärmen des kaltgewalzten Bandes von der Temperatur T1 auf eine Temperatur oberhalb (Ac3 - 20), vorzugsweise auf den Temperaturbereich von (Ac3 - 20) - (Ac3 + 20), besonders bevorzugt (Ac3 - 15) - (Ac3 + 15), mit einer Erwärmungsgeschwindigkeit V2 von 0,5 - 4,0 °C/s, vorzugsweise 1,0 - 3,0 °C/s, umfasst.

7. Verfahren nach einem der vorhergehenden Ansprüche, wobei in Schritt d) die der Wärmebehandlung im Bereich von Bn - (Ms + 50), vorzugsweise während einer Zeitspanne t5 von 40 - 100 Sekunden, durchgeführt wird.

8. Verfahren nach einem der vorhergehenden Ansprüche, umfassend einen weiteren Wärmebehandlungsschritt zwischen den Schritten d) und e) der Wärmebehandlung des aus Schritt c) resultierenden Stahlbandes im Bereich von Bs - Bn, vorzugsweise (Bs - 50) - Bn, vorzugsweise über eine Zeitdauer t6 von 5 - 30 Sekunden, stärker bevorzugt über eine Zeitdauer t6 von 10 - 20 Sekunden.

9. Verfahren nach Anspruch 8, wobei der weitere Wärmebehandlungsschritt eine Feuerverzinkungsbehandlung umfasst.

10. Verfahren nach einem der vorhergehenden Ansprüche 1 bis 8, das nach der Wärmebehandlung einen Beschichtungsschritt umfasst, bei dem das wärmebehandelte Stahlband mit einer Schutzschicht, vorzugsweise einer Zn-Schicht oder Zn-Legierungsschicht, beschichtet wird.

11. Verfahren nach einem der vorhergehenden Ansprüche, wobei die Mikrostruktur in Vol.-% Folgendes umfasst:

polygonalen Ferrit (PF):	0 - 5;
polygonalen Ferrit (PF) + nadelförmigen Ferrit (AF) + oberen bainitischen Ferrit (HBF):	10 - 25;
unteren bainitischen Ferrit (LBF) + partitionierten Martensit (PM):	55 - 80;
Restaustenit (RA):	7 - 15;
Martensit (M):	0 - 10;

und/oder wobei der C-Gehalt im Restaustenit (RA) 0,90 Gew.-% oder mehr, vorzugsweise 0,95 Gew.-% oder mehr beträgt.

12. Verfahren nach einem der vorhergehenden Ansprüche, wobei das resultierende Stahlband mindestens eine, vorzugsweise alle, der folgenden Eigenschaften aufweist:

Die Streckgrenze (YS) beträgt mindestens 550 MPa, und/oder
die Zugfestigkeit (TS) beträgt mindestens 980 MPa, und/oder
die Gesamtdehnung (TE) beträgt mindestens 13 %, und/oder
das Lochaufweitungsvermögen (HEC) beträgt mindestens 20 %, und/oder
der Biegewinkel (BA) beträgt mindestens 80°.

Revendications

1. Procédé de traitement thermique d'une bande d'acier laminée à froid, lequel procédé comprend les étapes de :

a) trempage d'une bande d'acier laminée à froid au-dessus de (Ac3-20) pendant un temps de trempage t2 de 1 à 200 secondes, obtenant ainsi une bande d'acier laminée à froid possédant une microstructure austénitique ;
b) refroidissement de la bande d'acier trempée résultant de l'étape a) jusqu'à une température T4 dans la plage Ms à (Ms - 200), comprenant une sous-étape de refroidissement de la bande d'acier trempée résultant de l'étape a) à une température T3 dans la plage de 800 à 550°C à un taux de refroidissement V3 de 2,0 à 15,0°C/s, et une sous-étape de refroidissement de la bande d'acier trempée à partir de la température T3 jusqu'à la température T4 à un taux de refroidissement V4 de 20,0 à 70,0°C/s ;
c) chauffage de la bande d'acier refroidie résultant de l'étape b) jusqu'à une plage de températures de Bs à Ms ;
d) traitement thermique de la bande d'acier chauffée dans la plage de températures de Bs à Ms pendant une durée t5 de 30 à 120 secondes ;
e) refroidissement de la bande d'acier traitée thermiquement à température ambiante ;

de sorte que la bande d'acier possède une microstructure (en % en volume) comprenant

ferrite polygonale (PF) :	0 - 10 ;
ferrite polygonale (PF) + ferrite aciculaire (AF) + ferrite bainitique supérieure (HBF) :	5 - 30
ferrite bainitique inférieure (LBF) + martensite partitionnée (PM) :	50 - 85 ;
austénite résiduelle (RA) :	5 - 20
martensite (M) :	0 - 15 ;

ladite bande d'acier possédant une composition (en pourcentage en masse) comprenant

C	0,15 - 0,28 ;
Mn	1,70 - 3,00 ;
Si	0,50 - 2,00 ;
Al	0,01 - 0,60 ;
P	inférieur à 0,050 ;
S	inférieur à 0,020 ;
N	inférieur à 0,0080 ;

ladite somme (Si + Al) étant ≥ 0,60 ; et
10C + Mn + Cr ≥ 3,85 et 8,5 ≤ (Mn + Cr)/C ≤ 16 ; éventuellement un ou plusieurs éléments choisis parmi

$$0 < Cr \leq 1{,}00 ;$$

$$0 < Cu \leq 0{,}20 ;$$

$$0 < Ni \leq 0{,}50 ;$$

$$0 < Mo \leq 0{,}50 ;$$

$$0 < Nb \leq 0{,}10 ;$$

$$0 < V \leq 0{,}10 ;$$

$$0 < Ti \leq 0{,}10 ;$$

$$0 < B \leq 0{,}0030 ;$$

$$0 < Ca \leq 0{,}0050 ;$$

$$0 < REM \leq 0{,}0100,$$

REM étant un ou plusieurs métaux des terres rares ; et le reste étant du fer et des impuretés inévitables.

2. Procédé selon la revendication 1, ladite étape c) impliquant le traitement thermique de la bande refroidie de l'étape b) à une température T4 dans la plage de températures de Ms à (Ms - 200), mieux encore dans la plage de températures de (Ms - 50) à (Ms - 150), de préférence ladite durée totale t4 de l'étape c) étant dans la plage de 1 à 10 secondes, mieux encore dans la plage de 1 à 5 secondes.

3. Procédé selon la revendication 1 ou la revendication 2, ladite étape a) comprenant le trempage d'une bande d'acier laminée à froid dans les limites d'une plage de températures de (Ac3 - 20) à (Ac3 + 20), de préférence dans les limites d'une plage de températures de (Ac3 - 15) à (Ac3 + 15), de préférence pour un temps de trempage t2 de 30 à 150 s.

4. Procédé selon l'une quelconque des revendications précédentes, ladite étape b) comprenant le refroidissement de la bande d'acier trempée de l'étape a) jusqu'à la température T4 à un taux de refroidissement suffisant pour éviter la formation de perlite.

5. Procédé selon l'une quelconque des revendications précédentes, ladite étape b) comprenant une sous-étape de refroidissement de la bande d'acier trempée résultant de l'étape a) jusqu'à une température T3 dans la plage de 750 à 600°C, de préférence à un taux de refroidissement V3 de 3,0 à 10,0°C/s.

6. Procédé selon l'une quelconque des revendications précédentes, avant l'étape a) comprenant en outre le chauffage d'une bande laminée à froid jusqu'à une température au-dessus de (Ac3 - 20) à un taux de chauffage supérieur ou égal à 0,5°C/s, comprenant de préférence le chauffage de la bande laminée à froid jusqu'à une température T1 dans la plage de 800 à 550°C, de préférence dans la plage de 750 à 600°C, à un taux de chauffage V1 de 10,0 à 30,0°C/s, de préférence à un taux de chauffage V1 de 15,0 à 25,0°C/s ; et en chauffant en outre la bande laminée à froid à partir de la température T1 jusqu'à une température au-dessus de (Ac3 - 20), de préférence jusqu'à la plage de températures de (Ac3 - 20) à (Ac3 + 20), mieux encore (Ac3 - 15) à (Ac3 + 15), à un taux de chauffage V2 de 0,5 à 4,0°C/s, de préférence de 1,0 à 3,0°C/s.

7. Procédé selon l'une quelconque des revendications précédentes, dans ladite étape d) ledit traitement thermique étant réalisé dans la plage de Bn à (Ms + 50), de préférence durant une période de temps t5 de 40 à 100 secondes.

8. Procédé selon l'une quelconque des revendications précédentes, comprenant une étape de traitement thermique supplémentaire entre les étapes d) et e) de traitement thermique de la bande d'acier résultant de l'étape c) dans la plage de Bs à Bn, de préférence (Bs - 50) à Bn, de préférence pendant une période de temps t6 de 5 à 30 secondes, mieux encore pendant une période de temps t6 de 10 à 20 secondes.

9. Procédé selon la revendication 8, ladite étape de traitement thermique supplémentaire comprenant un traitement de galvanisation à chaud au trempé.

10. Procédé selon l'une quelconque des revendications précédentes 1 à 8, après le traitement thermique, comprenant en outre une étape de revêtement pour revêtir la bande d'acier traitée thermiquement avec un revêtement protecteur, de préférence un revêtement de Zn ou d'alliage de Zn.

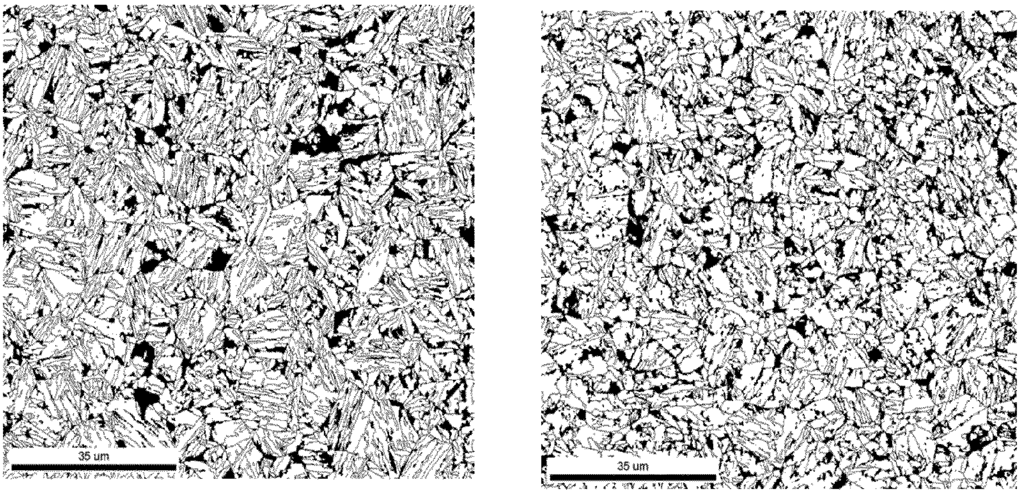
11. Procédé selon l'une quelconque des revendications précédentes, ladite microstructure comprenant en % en volume :

ferrite polygonale (PF) :	0 - 5 ;
ferrite polygonale (PF) + ferrite aciculaire (AF) + ferrite bainitique supérieure (HBF) :	10 - 25 ;
ferrite bainitique inférieure (LBF) + martensite partitionnée (PM) :	55 - 80 ;
austénite résiduelle (RA) :	7 - 15 ;
martensite (M) :	0 - 10 ;

et/ou ladite teneur en C dans l'austénite résiduelle (RA) étant de 0,90 % en poids ou plus, de préférence de 0,95 % en poids ou plus.

12. Procédé selon l'une quelconque des revendications précédentes, ladite bande d'acier résultante possédant au moins l'une, de préférence l'ensemble, des propriétés :

la limite d'élasticité (YS) est supérieure ou égale à 550 MPa ; et/ou
la résistance à la traction (TS) est supérieure ou égale à 980 MPa ; et/ou
l'allongement total (TE) est supérieur ou égal à 13 %, et/ou
la capacité d'extension de trous (HEC) est supérieure ou égale à 20 % ; et/ou
l'angle de pliage (BA) est supérieur ou égal à 80°.


35 um
35 um

(a) (b)

Fig. 1

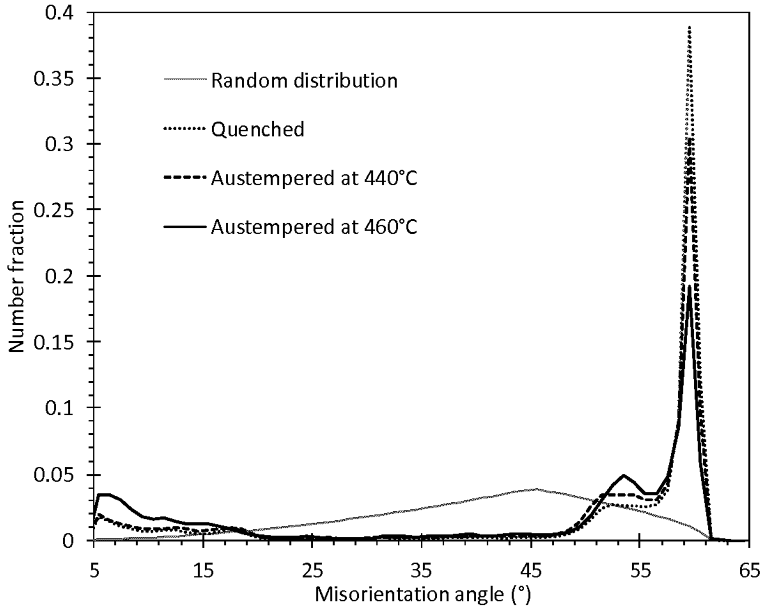

0.4
0.35
0.3
0.25
0.2
0.15
0.1
0.05
0
Number fraction
Random distribution
Quenched
Austempered at 440°C
Austempered at 460°C
5
15
25
35
45
55
65
Misorientation angle (°)

Fig. 2

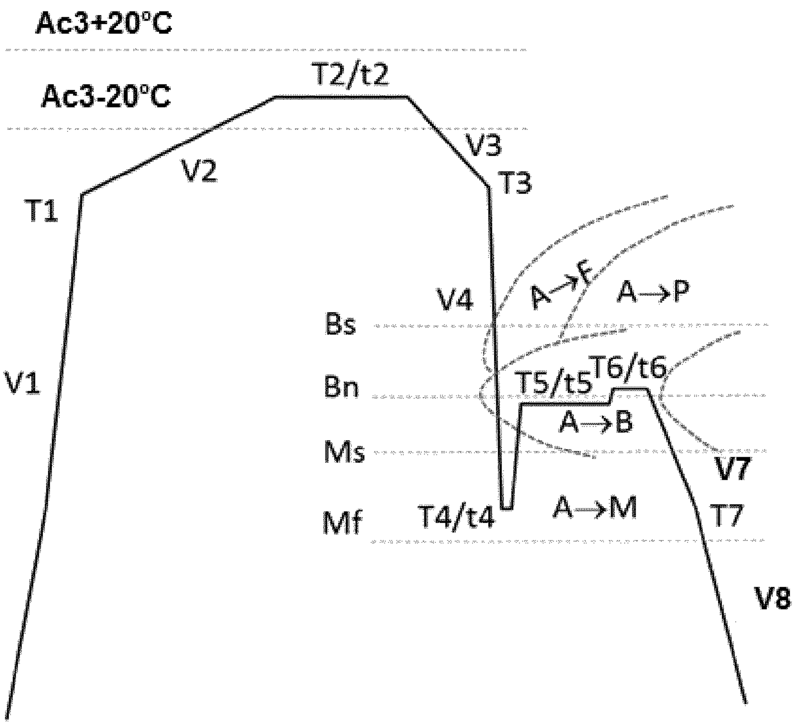

Ac3+20°C
Ac3-20°C
T2/t2
V3
T3
V2
T1
V4
A→F
A→P
Bs
V1
Bn
T5/t5
T6/t6
A→B
Ms
V7
Mf
T4/t4
A→M
T7
V8

Fig. 3

REFERENCES CITED IN THE DESCRIPTION

Patent documents cited in the description

- WO 2013144373 A1 **[0008]**
- EP 2831296 B1 **[0009]**
- EP 2831299 A **[0009]**
- WO 2018115936 A **[0009]**

Non-patent literature cited in the description

- **A.-F. GOURGUES** ; **H. M. FLOWER** ; **T. C. LINDLEY**. *Materials Science and Technology*, January 2000, vol. 16, 26-40 **[0066]**